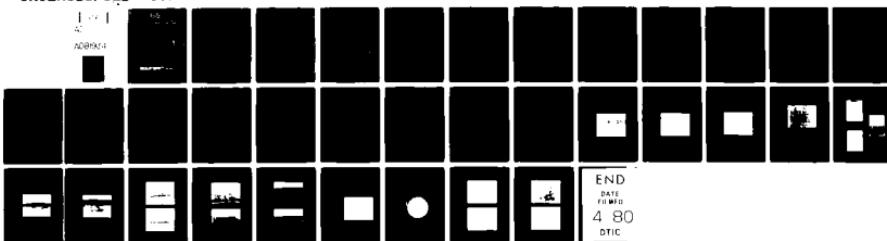


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COATINGS FOR DIRECTIONALLY SOLIDIFIED GAMMA PRIME-GAMMA PLUS ALPHA EUTECTICS

Final Report
October 1979

Prepared Under Contract N62269-78-C-0311

Naval Air Development Center
Warminster, PA 18974

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for

Naval Air Systems Command
Department of the Navy
Washington, DC 20361

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the coating. Coatings containing 37 w/o platinum did not crack in the 2100^oF cyclic oxidation tests.

GAMMA PRIME
ALPHA

Chromium used as an intervening layer between γ' and the overlay coating reacted with the substrate to form brittle phases which cracked during thermal cycling. The substitution of nickel for chromium as an intervening layer resulted in void formation (Kirkendall porosity) leading to eventual coating delamination.



FOREWORD

The work reported herein was performed by the United Technologies Research Center, East Hartford, Connecticut, 06108, for the Naval Air Development Center, Warminster, Pennsylvania, under Contract Number N62269-78-C-0311 during the period from August 7, 1978 to November 5, 1979. Mr. Irving Machlin of the Naval Air Systems Command, Washington, DC, was Project Technical Consultant.

Mr. Norman S. Bornstein, Chief, Materials Process Research, UTRC, was Program Manager, and Dr. John G. Smeggil was Principal Investigator.

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SECTION I.

INTRODUCTION

A. General Background

1. Requirements

The directionally solidified eutectic alloys as represented by gamma/gamma prime plus alpha, $\gamma/\gamma'-\alpha$ (Ni-31Mo-6.2Al), and gamma prime-alpha, $\gamma'-\alpha$ (Ni-18.3 Mo-16.6Ta-6.8Al), are attractive advanced alloys which offer a 93°C (200°F) metal temperature advantage or a 50 percent strength increase over the best of the current nickel-based superalloys. These alloys are being developed for use in advanced Naval aircraft gas turbine engines.

The oxidation/corrosion characteristics of the directionally solidified eutectics have been studied, and coatings will be required to protect the substrate alloys over a wide range of temperatures (Refs. 1 and 2). Accordingly, external airfoil surfaces, internal cooling channels, and platform sections will need to be protected. The cooler root sections may or may not require a coating; the decision will depend upon the final alloy composition (Ref. 3).

The goal of this program is to develop and evaluate oxidation and hot corrosion resistant coatings for directionally solidified $\gamma/\gamma'-\alpha$ and $\gamma'-\alpha$ alloys for use in advanced Naval aircraft gas turbines.

2. Coating Processes

a. Reaction Coatings

The earliest coatings used to extend the useful life of gas turbine superalloys are the "simple" aluminides, which are approximately 0.0015 to 0.003 inch (1.5 to 3 mils) thick. The aluminide coatings can be formed by applying aluminum onto airfoil surfaces and subsequently heat treating at elevated temperatures or alternatively by packing the component in a canister containing aluminum, an activator (usually a halide salt such as ammonium chloride), and an inert dispersant (to prevent pack sintering) and again heat treating at elevated temperatures. In these processes aluminum reacts with and diffuses into the superalloy substrate to form the coating which is composed primarily of the intermetallic phase NiAl (or CoAl) but also contains all of the substrate alloying elements. The elements which are insoluble in the aluminide coating precipitate, and these precipitated phases can degrade the oxidation/hot corrosion performance of the coatings.

Molybdenum is highly insoluble in NiAl, and its oxide which readily forms at elevated temperatures reacts with and fluxes the normally protective alumina scales thereby markedly increasing the oxidation rate of the coating.

The effect of alloy substrate chemistry with respect to coating composition can be minimized by forming the coating by the outward diffusion of nickel rather than the inward diffusion of aluminum. This is accomplished by limiting the activity of aluminum in the pac' usually by the use of prealloyed powders. Dimensional changes which are related to the differences in specific volumes of the reactive phases are minimal, as is the case for the coatings formed by inward diffusion. The coatings formed by the outward diffusion of the principal substrate component are single phase, saturated with the matrix alloying elements. A major difficulty of the coating, however, is the tendency towards the formation of pores or holes which will accumulate at or near the coating-substrate interface; the formation of these defects are related to the well known Kirkendall effect.

The addition of precious metals such as platinum, palladium, or rhodium to the aluminide coatings have resulted in increased oxidation/corrosion resistance although the mechanism by which the improvement occurs is not fully understood. It is believed that these precious metal additions increase the adherence of the protective alumina scale which forms during service. However, as previously discussed, the relatively uniform distribution of the corrosion prone, insoluble, refractory metal-precipitated phases tends to negate any beneficial effect of the precious metal alloy addition.

b. Overlay Coatings

"Overlay" is the general term used to describe coatings whose initial composition is independent of substrate chemistry.

In the physical vapor deposition (PVD) processes the part to be coated is exposed to the metallic vapors generated by the electron beam heating of an ingot of the desired composition. The process is conducted in a vacuum chamber. An alternate process involves sputtering where the atoms of a selected target are accelerated to the surface of the part to be coated.

Conventional or RF diode sputtering transfers material from a liquid (evaporation) or solid (sublimation) state source. In the triode sputtering system also used in this study, a tungsten wire, heated to a sufficiently high temperature, thermionically emits electrons. These electrons are weakly accelerated to the opposite anode by a small positive potential difference potential. In transit through the atmosphere between the tungsten wire filament and the positive anode, a percentage of the electrons collide with some of the argon atoms yielding two electrons from the original atom plus one electron. The argon ions near the target (the coating material source) are accelerated toward it by an impressed negative potential. Such argon ions impact the target surface, and a percentage of them eject atoms from the target. These ejected target atoms traverse the open region, condensing onto the substrate to form a coating.

The dc triode sputtering system is used primarily to effect advanced oxidation-hot corrosion resistant and wear resistant coatings. Multi-element coatings are deposited at controlled rates from alloy targets onto heated rotating substrates. Complex alloy systems, including 10 element compositions, have been successfully deposited by such sputtering.

3. Problem and Approach

The major problem encountered in the coating of the directionally solidified eutectic alloys are related to:

- coating-substrate interdiffusion and
- thermomechanical coating-substrate interactions.

During cyclic exposure the current family of protective overlay MCrAlY, M = Co and/or Ni, coatings tend to crack, due to the stresses which develop as a result of the differences in coefficients of thermal expansion between the substrate and the coating. Moreover, the mechanical properties of directionally solidified eutectic alloys in large part derive from their aligned microstructures (Ref. 1, 2, and 4). However, protective coating-eutectic substrate interdiffusion effects can markedly alter such eutectic microstructures, thereby adversely affecting mechanical properties (Ref. 4).

In the previous study it was shown that although the coefficient of thermal expansion of the beta phase (NiAl) closely matches that of the directionally solidified eutectic alloys, Tables I and II, the coating is not diffusionally stable and, during prolonged exposures at elevated temperatures, diffuses into the substrate (Ref. 5). Although chromium additions to the coatings did reduce the tendency of the overlay coatings to crack, it was concluded that the chromium content of the coating cannot be sufficiently increased to obtain both low temperature crack free coatings and high temperature diffusional stability. The approach that was shown to be most promising was that of increasing the strain capacity of the coating, which is accomplished by the addition of platinum. Platinum also reduces the average coefficient of thermal expansion of the coating. Therefore, an objective of this study is to demonstrate that in dynamic burner rig tests the platinum modified NiCoCrAlY coating will protect the γ/γ' - α substrate during high temperature oxidation and hot corrosion tests.

The family of directionally solidified eutectic alloys as represented by γ' - α are anticipated to be more difficult to protect since the gamma phase is removed from the alloy. The difficulties anticipated are based not only on compositional changes, but thermomechanical differences since the deletion of the gamma phase would tend to further reduce the average coefficient of thermal expansion of the alloy. For example, in the temperature range 100-1000°C, the average coefficient of thermal expansion for Ni₃Al (γ') is 13.4°C⁻¹ while that for gamma (γ) is 17.1°C⁻¹ (Ref. 6). Thus in this study the approach taken for the development of the platinum modified NiCoCrAlY coating (for γ/γ' - α) will be applied toward the γ' - α alloys with emphasis directed towards improved strain capacity and the use of diffusion barriers.

SECTION II

EXPERIMENTAL PROCEDURES

A. Eutectic Substrate Preparation

The alloys employed in this study consisted of the gamma/gamma prime plus alpha alloy of the nominal composition Ni-33 w/o Mo-5.5 w/o Al and the gamma prime-alpha alloy of the nominal composition Ni-18.3 w/o Mo 16.6 w/o Ta 6.8 w/o Al. All specimens were fabricated from 1 $\frac{1}{4}$ " diameter bars directionally solidified at a rate of 1.5cm/hour. No cellular microstructural features were observed in any of the longitudinal stripes polished into the prepared ingots. The as processed bars were used in the burner rig tests, and laboratory specimens approximately 0.2cm thick were cut and polished to 600 grit SiC for all static tests.

B. Coating Deposition

The burner rig test specimens were coated with a platinum-modified NiCoCrAlY composition previously identified for this study, i.e., the platinum content of the modified NiCoCrAlY was at least 15 wt.-% (Ref. 5) . Prior to coating, all specimens were cleaned as follows:

1. a light blast of Novaculite 200 (~40 psi at 6"),
2. a chlorinated abrasive scrub with hot water,
3. a cold water wash/rinse,
4. an ethanol rinse,
5. dry with an argon blast.

The coatings were applied in triode sputtering system. Based on standardization runs, time was used to control deposit thickness. A supplemental target was co-sputtered with the main target in order to add platinum to the coating. The rate of platinum deposition was controlled by the voltage impressed on the supplemental target. In all of the sputtering processes the specimens were slowly rotated (2 RPM) during the coating process to insure an even deposit.

The laboratory test specimens were primarily coated using the RF diode sputtering apparatus, although some specimens were coated in the triode system. The choice of systems depended upon availability.

The cleaning procedure was used in RF sputtering was analogous to that used in the triode system, differing only in the substitution of trichloroethylene for the Novaculite 200 (Step 1), a methanol rinse for an ethanol rinse (Step 4), and a nitrogen blast rather than an argon blase (Step 3). For support a 30 mil diameter platinum wire was attached to the substrate by tack welding. The power used for this welding was approximately 95 watt-seconds. The vacuum systems

were all pumped down into the low 10^{-7} torr range prior to deposition. After deposition all specimens were heated for four hours at 1975°F (1079°C) in flowing argon. Compositions of sputtered deposits were determined by electron microprobe techniques.

C. Oxidation/Hot Corrosion Testing

1. Dynamic Burner Rig Studies

The conventional Mach 0.3 burner rig was used to determine the oxidation/hot corrosion behavior of the NiCoCrAlY-coated gamma/gamma prime plus alpha ($\gamma/\gamma'-\alpha$) specimens. The fuel in all tests was Jet A. The specimen temperatures were measured directly with a Leeds and Northup optical pyrometer Model No. 8622 and indirectly by thermocouples sensing the hot gas stream near the test specimens. In all tests the specimens are rotated in a carousel at 1750 RPM. A hot corrosion environment is established by the introduction of the standard aqueous synthetic sea salt solution into the burner near the swirler plate. The concentration of the salt in the environment was 10 ppm.

2. Laboratory Experiments

In the cyclic laboratory oxidation tests, the specimens are exposed at 2100°F (1150°C) for 55 minutes then cooled to room temperature for 5 minutes. The time required for the specimens to go from 2100°F to about 210°F (100°C) is approximately 15 seconds.

The cyclic hot corrosion test consisted of exposing the $1 \text{ mg/cm}^2 \text{ Na}_2\text{SO}_4$ coated specimens for 55 minutes at 900°C and 5 minutes at 200°C. In general, it took about 1 minute for the hot specimen to cool to the ambient temperature. The hot corrosion environment was continuously maintained by recoating the specimens every 24 hours with a fresh application of sodium sulfate.

D. Thermal Expansion Experiments

A Theta II Research dilatometer was used in conjunction with an NBS platinum standard to provide the thermal expansion data which was performed in an argon atmosphere in the temperature range from room temperature to 1300°C.

SECTION III

RESULTS AND DISCUSSION

A. Dynamic Burner Rig Studies

The platinum modified overlay coating which was previously identified for this study contained a minimum of 15 weight percent of the precious metal. All of the coatings applied by the sputtering technique previously described were characterized by quantitatively chemically analyzing test coupons and also by determining the chemical composition of the coating during routine electron microprobe studies. The nominal composition of the platinum modified NiCoCrAlY coating is as follows: 43.3 w/o Ni, 13.8 w/o Co, 15.8 w/o Cr, 9.2 w/o Al, 16.9 w/o Pt, and 0.1 w/o Y (Ref. 7). The average coating thickness is five (5) mils.

In the dynamic burner rig tests the platinum modified NiCoCrAlY coating quickly formed an adherent oxide layer which effectively separates the hostile oxidizing environment from the coating substrate. In all of the one hundred hour tests, the weight changes were for all practical purposes negligible, in agreement with the metallographic studies which are summarized in Figures 1, 2, and 3, cf. Section II C1.

The microstructure of the as received coating consists of an inner diffusion layer, approximately 1 mil thick, which is formed during the 4 hour 1975°F diffusion heat treatment. Even after 100 hours exposure at 1650°F, no apparent growth of either the outward Pt modified NiCoCrAlY coating or the diffusion layer are observed, Figure 1. Moreover, the thickness of the adherent oxide scale is too small to resolve by common optical (light) metallography. Lastly, it should be noted that the original coating interface is inadvertently defined by the presence of inert oxide particles, which were not removed during the cleaning process and were trapped during coating. The thickness of the diffusion zone is approximately one mil, and the molybdenum associated with this volume of matrix is in solution within the coating; the majority partitioning to the terminal solid solution phase since molybdenum is relatively insoluble in the beta phase.

At 1800°F the results are for all practical purposes identical to those reported for the 1650°F tests, in that no morphological changes were observed with respect to either the coating or the matrix, Figure 2. Even after 100 hours, no significant weight changes were noted.

At 2100°F the platinum modified NiCoCrAlY again formed an adherent protective oxide scale, and although the weight changes were minimal, significant interdiffusional effects were noted. The microstructure of the coating after 100 test hours is shown in Figure 3. The coating has increased in thickness approximately two fold, with both nickel and molybdenum diffusing from the substrate

into the coating. The composition of the phases in the coating and in the alloy-affected zone are shown in Figure 4. The molybdenum content of the solid solution phase is approximately 15 w/o, and approximately 6 w/o in the aluminum rich phase. It is also noted that the platinum is partitioned to the aluminum rich phase, 12 w/o as compared to 8 w/o in the matrix phase. Lastly, the distribution of aluminum, platinum, and molybdenum within the coating are shown in the x-ray maps also presented in Figure 4. It can be seen that platinum is diffusing into the eutectic alloy substrate, and molybdenum is entering the coating. The high concentration of aluminum at the coating surface is due to the alumina scale that protects the alloy.

The addition of ten parts per million of sea salt to the burner environment resulted in the deposition of salt onto the test specimens. In general, the butt section, the end held in the burner carousel, is uncoated. Thusly due to the very rapid rates of oxidation associated with the sulfidation attack that occurred in this cooler region, weight change measurements were meaningless.

The microstructure associated with hot corrosion tests for the specimens exposed at 1650°F and 1800°F are shown in Figures 5 and 6, respectively. In both figures it is apparent that no hot corrosion has occurred and that metallographically the coatings are identical in morphology to those exposed to the salt free environment as would be expected as long as the coating remained intact. At 2100°F, due to sulfidation attack of the uncoated butt section and the thinly coated tip section, catastrophic corrosion via the well-described "cut edge" effect occurred, resulting in catastrophic attack of the coating (Ref. 80. The "cut edge" effect involves preferential attack at the coating-substrate-condensed salt interface and is caused by molten metal sulfides present there (Ref. 8).

B. Laboratory Tests

It was anticipated that the average coefficient of thermal expansion of the γ' - α alloys would be less than that for the γ/γ' - α alloys because the former does not contain gamma, the phase with the larger coefficient of thermal expansion. The measured longitudinal thermal expansion coefficients for the γ' - α alloy in the temperature range from 100 to 1000°C are presented in Table III along with those of the γ/γ' - α alloy. Although as shown in Table III, the average values are quite close, the overall value for γ' - α is less than that for γ/γ' - α and this difference could influence coating performance.

In the prior study it was established that platinum-modified MCrAlY compositions produced adherent coatings, and in the current study it was established that the overlay coating containing at least 15 w/o Pt did not crack in burner rig tests. In the current laboratory tests, 18 w/o Pt was added to the basic NiCoCrAlY (Ni-23Co-18Cr-12Al-0.5Y) coating which was applied onto γ' - α substrates and tested under cyclic conditions at 1650°F (sulfidation) and 2100°F (oxidation) cf. Section II C2. The coating cracked along longitudinal substrate surfaces (Figure 7); however, no cracks were observed on transverse surfaces. Similar behavior was observed for specimens cyclically tested (300 cycles) at 2100°F, Figure 8.

Based upon our previous experience the addition of platinum increases both the strain capacity of the coating as well as reduces the average coefficient of thermal expansion, the concentration of platinum in the base line NiCoCrAlY coating was increased from 18 to 37 w/o. As shown in Figure 9 both at 1650°F and 2100°F, the coating did not crack during the thermal cycling, but at the higher temperature did diffuse into the substrate. The depth of diffusion at 2100°F was for all practical purposes identical to that observed for the $\gamma/\gamma'-\alpha$ alloys.

The use of chromium and nickel to form an intervening layer between the overlay coating and the substrate was also investigated. Thick (0.4 mil) and thin (0.2 mil) chromium layers were applied onto the $\gamma'-\alpha$ substrate, and the treated substrate was then covered with 37 w/o Pt-modified NiCoCrAlY overlay coating. The coating formed on the $\gamma'-\alpha$ matrix is shown in Figure 10. Upon heat treatment (4 hours, 1975°F argon atmosphere) the coating cracked, Figure 11; the cracks apparently originating at the chromium eutectic alloy interface. Based upon metallographic studies it is hypothesized that the chromium reacted with and modified the alpha molybdenum phase which contributed to the coating instability.

When the chromium was replaced with nickel, the coatings did not crack during the standard 4 hour-1975°F (argon atmosphere) diffusion heat treatment step. However, the nickel preferentially diffused into the matrix. The preferential diffusion of nickel into the matrix results in the formation of vacancies which tend to coalesce at the coating-substrate interface as shown in Figure 12. During 1650°F sulfidation tests and during 2100°F oxidation tests, the coating would then fail by delamination caused by void accumulation at this interface. In the sulfidation tests the coating was attacked by a "cut edge" effect as shown in Figure 13 in which the fused salt reacts with the substrate and simultaneously with the coating at the substrate-coating interface. Such "cut edge" attack is a normal mode of hot corrosion propagation once the protective coating has been penetrated and does not necessarily imply inadequate coating processing in such a location (Ref. 8).

SECTION IV

CONCLUSIONS

- The addition of platinum to the NiCoCrAlY system increases the strain capacity of the coating and reduces the average coefficient of thermal expansion of the coating.
- A minimum of approximately 17 w/o platinum to the NiCoCrAlY system (Ni-13.8 w/o Co-15.8 w/o Cr-8.2 w/o Al-16.9 w/o Pt-0.1 w/o Y) is sufficient to produce overlay coatings capable of protecting the directionally solidified eutectic alloy $\gamma/\gamma'-\alpha$, (Ni-31Mo-6.2Al), during cyclic oxidation and sulfidation tests and burner rig tests without forming cracks.
- In burner rig testing involving hot corrosion conditions at 2100°F, the platinum-modified overlay coating on $\gamma/\gamma'-\alpha$ failed due to preferential oxidation/hot corrosion of the substrate alloy; the primary mode of failure was "cut edge" attack.
- The average coefficient of thermal expansion of the $\gamma'-\alpha$ alloy (Ni-18.3Mo-16.6Ta-6.8Al) in the temperature range 100 to 1000°C is $11.0 \times 10^{-6} \text{C}^{-1}$ along longitudinal directions while that for the $\gamma/\gamma'-\alpha$ alloy is $11.4 \times 10^{-6} \text{C}^{-1}$.
- The addition of 18 w/o platinum to the NiCoCrAlY coating did not prevent crack formation from occurring on the overlay-coated $\gamma'-\alpha$ alloy.
- Overlay coatings containing 37 w/o platinum on $\gamma'-\alpha$ substrates did not crack during cyclic 2100°F oxidation tests or the 1650°F sulfidation tests.
- Chromium used as an intervening layer between $\gamma'-\alpha$ substrate and the overlay coating reacts with the substrate, modifying the molybdenum rich phase and forming phases which cracked during thermal cycling.
- The substitution of nickel as an intervening layer for chromium resulted in the formation of a zone of voids, the Kirkendall porosity occurring as a result of preferential diffusion of nickel into the substrate.
- In general in this study it was shown that platinum reduces the average coefficient of thermal expansion of overlay coatings and also increases the strain capacity of the coating. The concentration of platinum necessary to ensure coating-substrate compatibility is dependent upon substrate chemistry. Thus if a composition of either $\gamma/\gamma'-\alpha$ or $\gamma'-\alpha$ is selected for further evaluation as a potential high temperature structural material, then further work may be necessary to optimize the coating composition to such an alloy.

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- Although platinum-modified NiCoCrAlY coatings have been successfully tested on γ/γ' - α and γ' - α substrates, platinum is becoming increasingly more expensive. Thus an economic incentive exists to find an alternative coating formulation for the γ/γ' - α and γ' - α alloys.

SECTION V

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TABLE I

Average Coefficient of Thermal Expansion
for $\gamma/\gamma' - \alpha$ as a Function of Temperature

(Ni-31.5Mo-6.2Al)

$^{\circ}\text{C}$	Transverse	Longitudinal	Diagonal (45 $^{\circ}$)
	Average $10^{-6} \text{ }^{\circ}\text{C}^{-1}$	Average $10^{-6} \text{ }^{\circ}\text{C}^{-1}$	Average
100 - 200	11.2	8.2	9.6
200 - 300	11.5	9.6	10.6
300 - 400	12.0	10.5	11.2
400 - 500	12.5	11.3	12.0
500 - 600	12.0	11.4	12.0
600 - 700	13.7	11.6	12.6
700 - 800	16.0	12.3	14.8
800 - 900	16.5	13.0	15.8
900 - 1000	<u>19.3</u>	<u>14.5</u>	<u>16.3</u>
Mean Value (100-1000 $^{\circ}\text{C}$)	13.9	11.4	12.8

TABLE II
Average Coefficient of Thermal Expansion

<u>Temperature Interval</u> °C	<u>Substrate</u>				
	Ni ₃ Al*	NiAl*	CoCrAlY* (expressed in units of 10 ⁻⁶ °C ⁻¹)	NiCoCrAlY	NiCoCrAlY+Pt
100 - 200	11.3	11.7	10.7	10.5	9.3
200 - 300	12.2	12.8	12.0	12.4	11.2
300 - 400	12.9	13.2	12.6	13.1	11.8
400 - 500	13.3	13.6	13.2	14.0	12.4
500 - 600	13.7	13.9	13.5	14.5	13.0
600 - 700	14.0	14.0	14.2	15.4	13.6
700 - 800	14.3	14.0	14.8	16.3	14.4
800 - 900	14.5	14.1	15.5	17.5	15.5
900 - 1000	14.8	14.5	16.1	18.1	16.7
1000 - 1100	15.3	14.4	17.4	18.6	17.5
1100 - 1200	15.8	14.4	----	----	----
Mean Value					
Over Temperature Range Measured:					
	13.8	13.7	14.0	15.1	13.5

* Averaged values for multiple data.

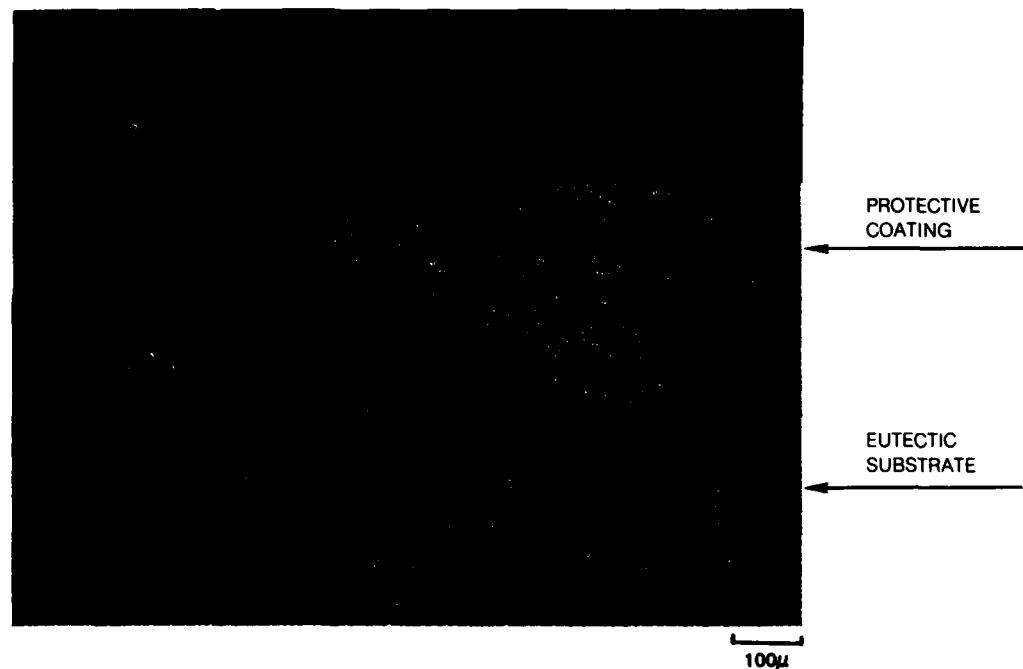
TABLE III

Average Coefficients of Thermal Expansion for the
Longitudinal Eutectic Direction as a Function of Temperature

$^{\circ}\text{C}$	$\gamma'/\gamma-\alpha$ (Ni-31.5Mo-6.2Al) ($10^{-6} \text{ }^{\circ}\text{C}^{-1}$)	$\gamma'-\alpha$ (Ni-18.3Mo-16.6Ta-6.8Al) ($10^{-6} \text{ }^{\circ}\text{C}^{-1}$)
100 - 200	8.2	9.8
200 - 300	9.6	10.0
300 - 400	10.5	10.4
400 - 500	11.3	10.8
500 - 600	11.4	10.9
600 - 700	11.6	11.0
700 - 800	12.3	11.4
800 - 900	13.0	12.1
900 - 1000	<u>14.5</u>	<u>13.8</u>
Mean Value (100-1000 $^{\circ}\text{C}$)	11.4	11.0

DYNAMIC BURNER RIG STUDIES

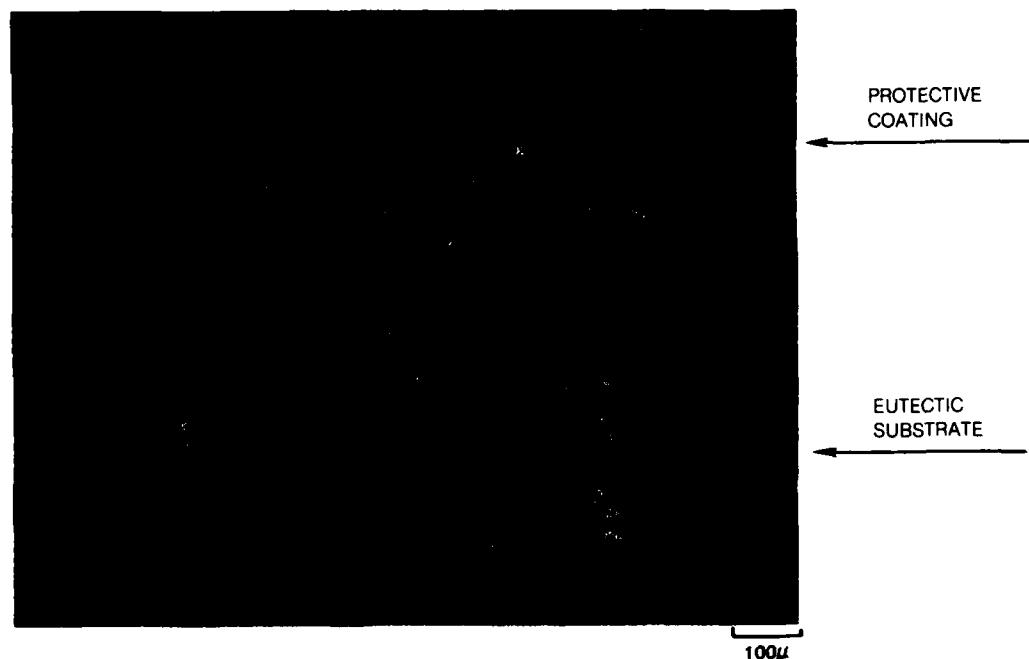
1650°F — 100 HOURS EXPOSURE
OXIDATION STUDIES



TYPICAL MICROSTRUCTURE OF Pt - MODIFIED OVERLAY
COATING ON GAMMA-GAMMA PRIME PLUS ALPHA SUBSTRATE

DYNAMIC BURNER RIG STUDIES

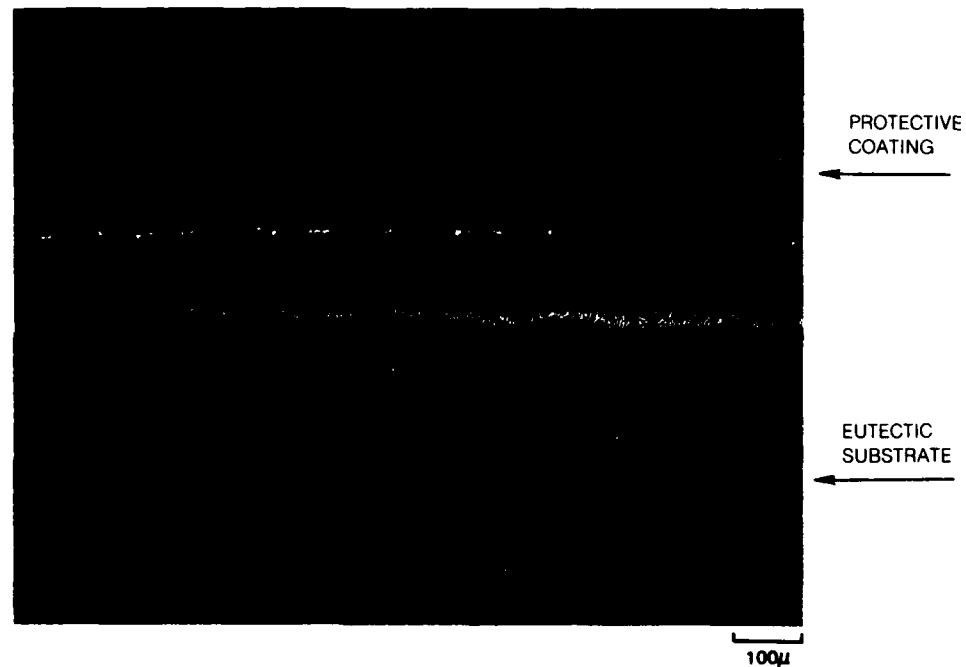
1800°F — 100 HOURS EXPOSURE
OXIDATION STUDIES



TYPICAL MICROSTRUCTURE OF Pt - MODIFIED OVERLAY
COATING ON GAMMA-GAMMA PRIME PLUS ALPHA SUBSTRATE

DYNAMIC BURNER RIG STUDIES

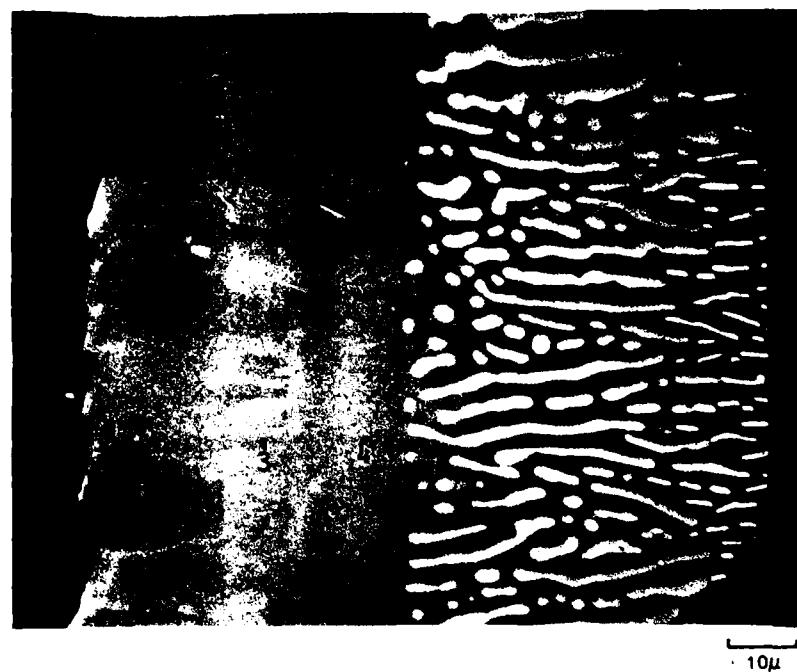
2100°F — 100 HOURS EXPOSURE
OXIDATION STUDIES



TYPICAL MICROSTRUCTURE OF Pt - MODIFIED OVERLAY
COATING ON GAMMA-GAMMA PRIME PLUS ALPHA SUBSTRATE

FIG. 4

ELECTRON MICROBEAM STUDIES OF PLATINUM-MODIFIED OVERLAY COATING



POSITION	CHEMICAL COMPOSITION W/O					
	Cr	Al	Co	Pt	Mo	Ni*
1	4.1	6.6	3.8	8.7	14.8	61.7
2	1.6	12.2	2.3	12.3	6.3	65.2
3	4.0	7.1	3.8	8.1	16.2	60.7
4	3.9	7.2	3.6	6.4	17.0	61.9

• BY DIFFERENCE

FIG. 4 (CONT'D)

ELECTRON MICROBEAM STUDIES OF PLATINUM-MODIFIED OVERLAY COATING
X-RAY MAPS

ALUMINUM



MOLYBDENUM

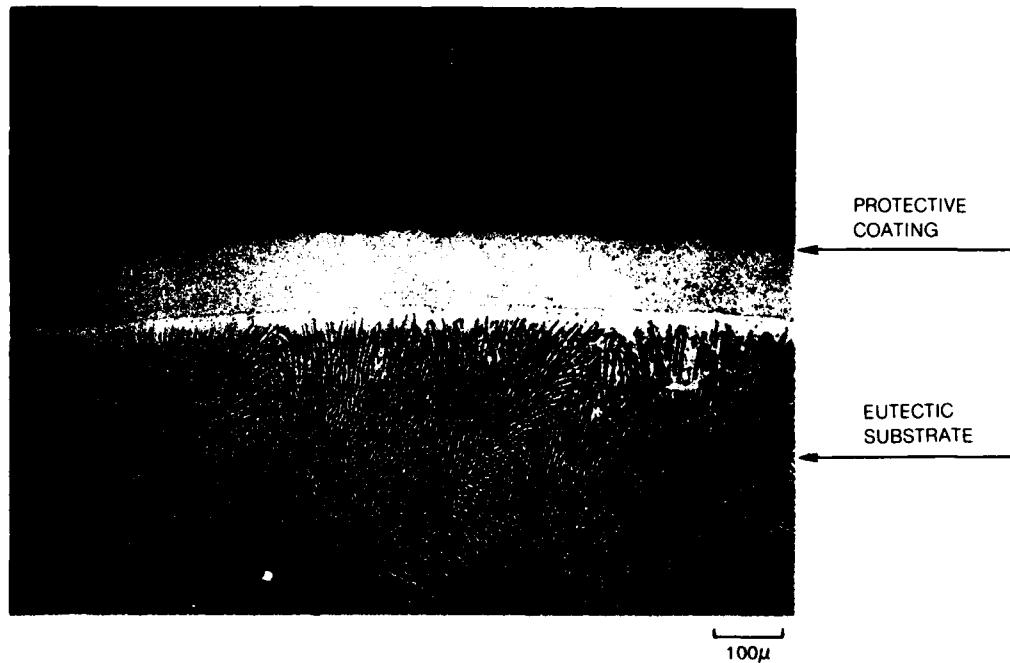


PLATINUM



DYNAMIC BURNER RIG STUDIES

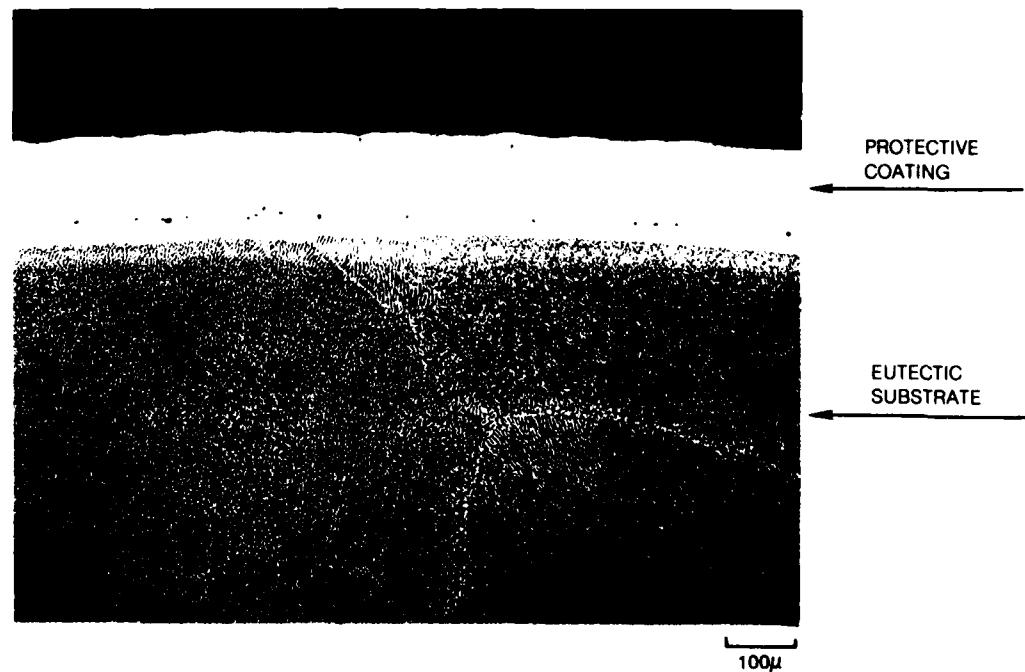
1650°F — 100 HOURS EXPOSURE
HOT CORROSION STUDIES



TYPICAL MICROSTRUCTURE OF PI-MODIFIED
OVERLAY COATING ON GAMMA/GAMMA
PRIME PLUS ALPHA STRUCTURE

DYNAMIC BURNER RIG STUDIES

1800°F — 100 HOURS EXPOSURE
HOT CORROSION STUDIES

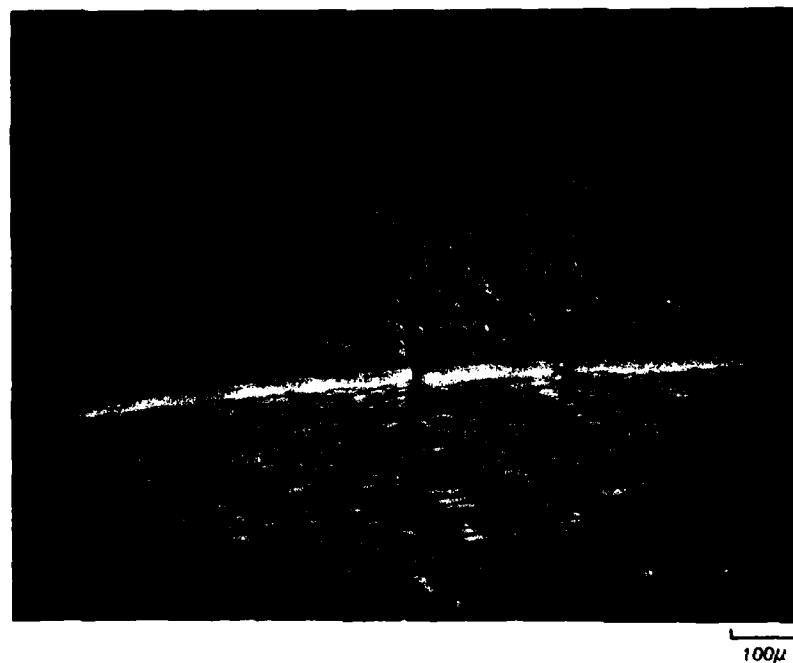


TYPICAL MICROSTRUCTURE OF PI-MODIFIED
OVERLAY COATING ON GAMMA/GAMMA
PRIME PLUS ALPHA SUBSTRATE

**CYCLIC SULFIDATION STUDIES OF PLATINUM-MODIFIED
OVERLAY COATINGS ON GAMMA PRIME-ALPHA**

300 CYCLES (1 HOUR/CYCLE)

LONGITUDINAL SURFACE



TRANSVERSE SURFACE

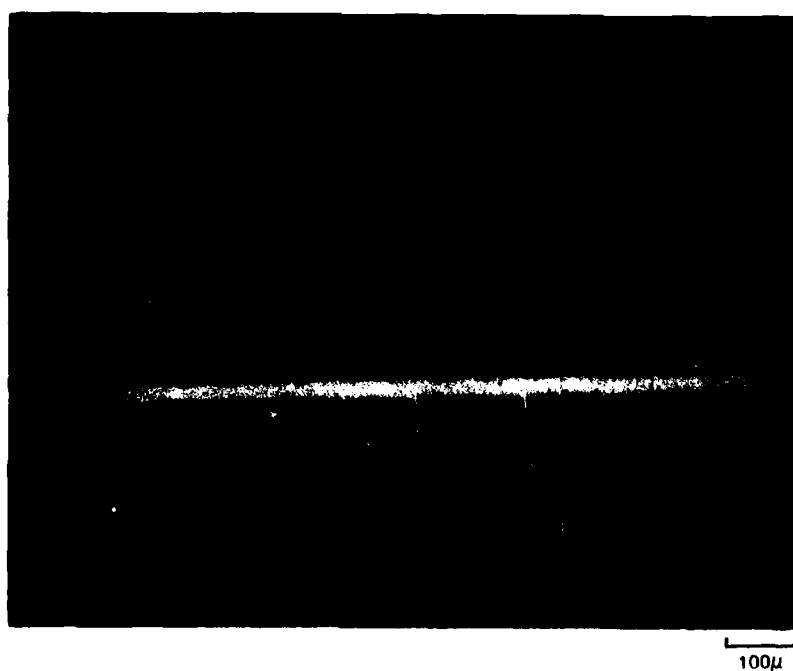
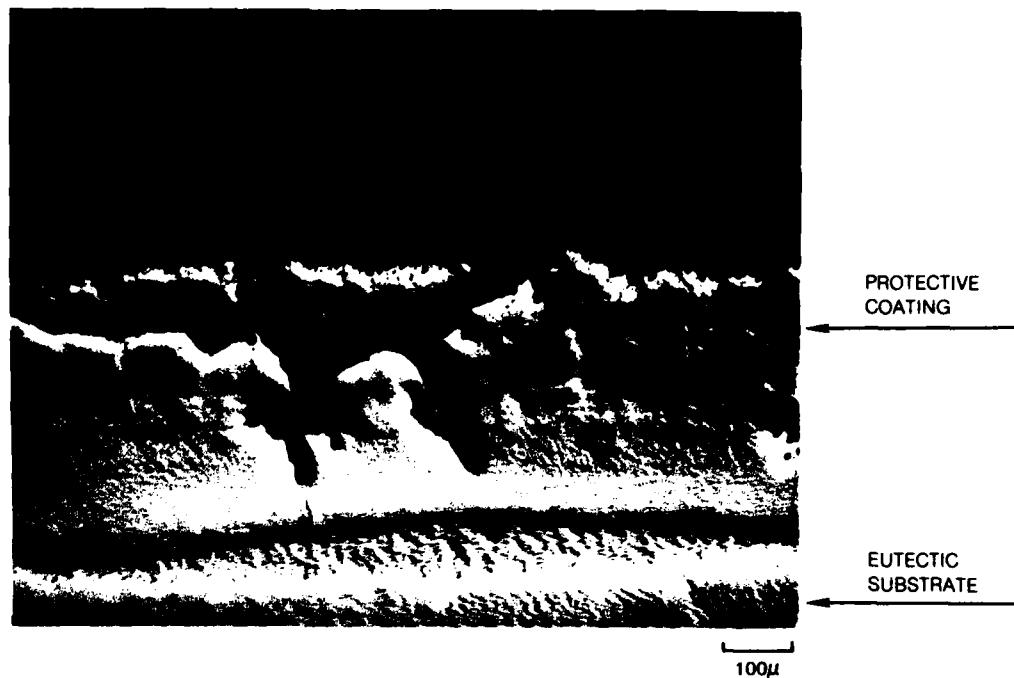


FIG. 8

**CYCLIC OXIDATION STUDIES OF PLATINUM-MODIFIED
OVERLAY COATINGS ON GAMMA PRIME-ALPHA**

300 CYCLES (1 HOUR/CYCLE)

LONGITUDINAL SURFACE



TRANSVERSE SURFACE

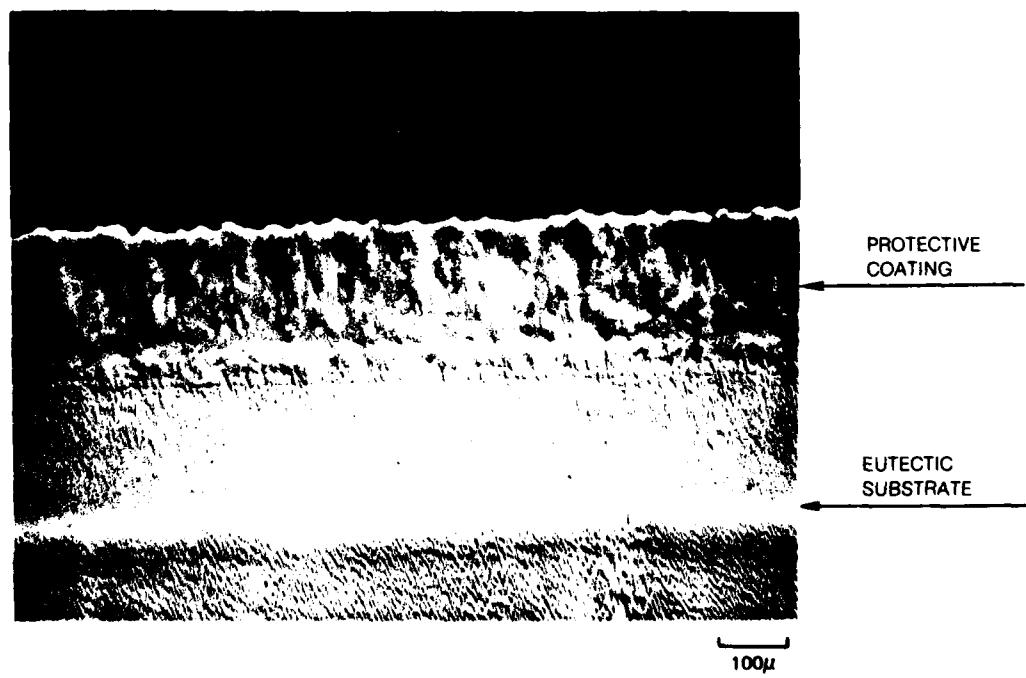
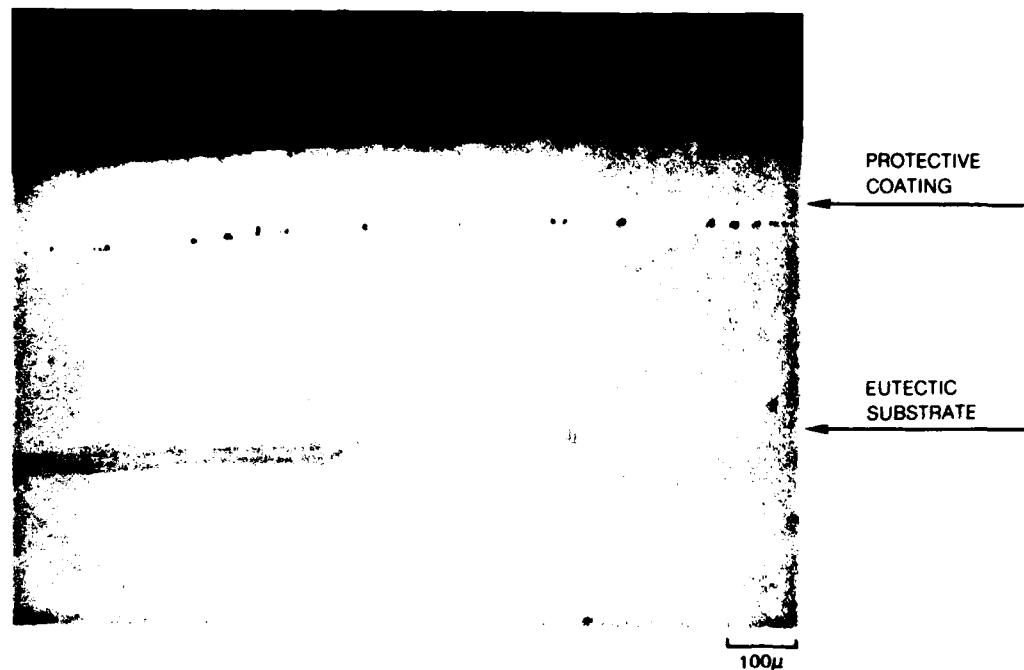


FIG. 9

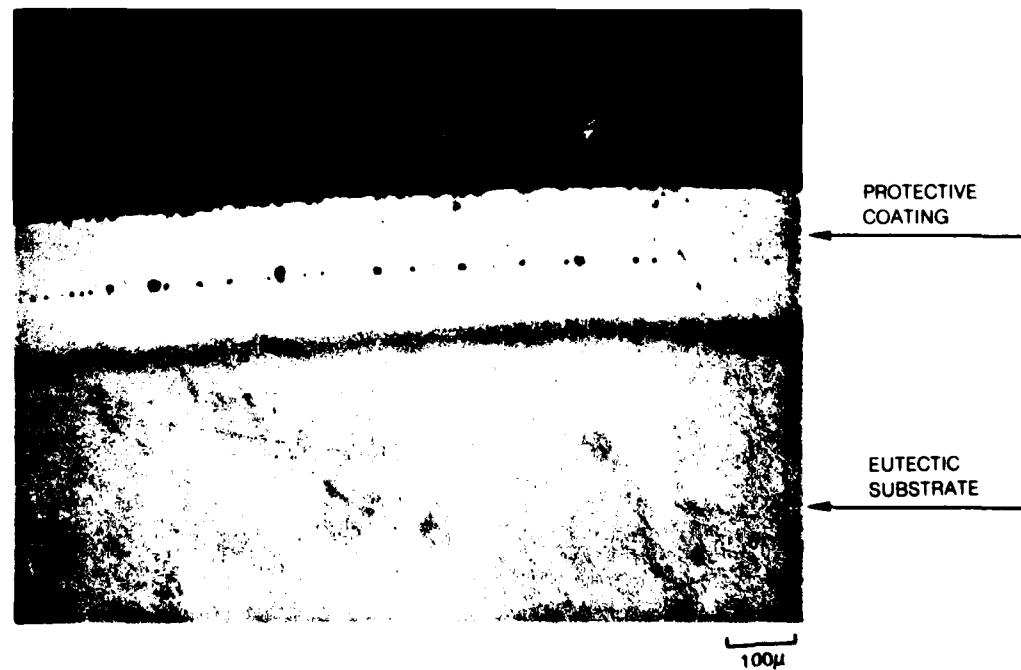
**CYCLIC STUDIES OF 37 WT% PLATINUM-MODIFIED
OVERLAY ON GAMMA PRIME-ALPHA ALLOYS**

300 CYCLES (1 HOUR/CYCLE)

SULFIDATION CORROSION 1650°F



OXIDATION 2100°F



STUDIES INVOLVING INTERVENING CHROMIUM LAYERS

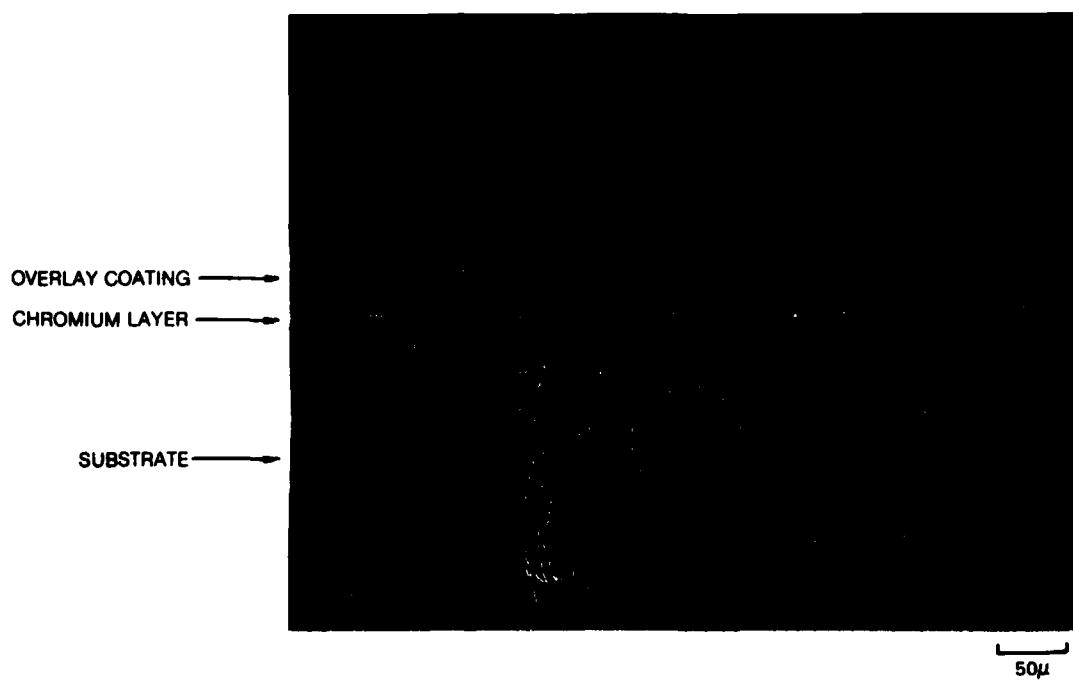
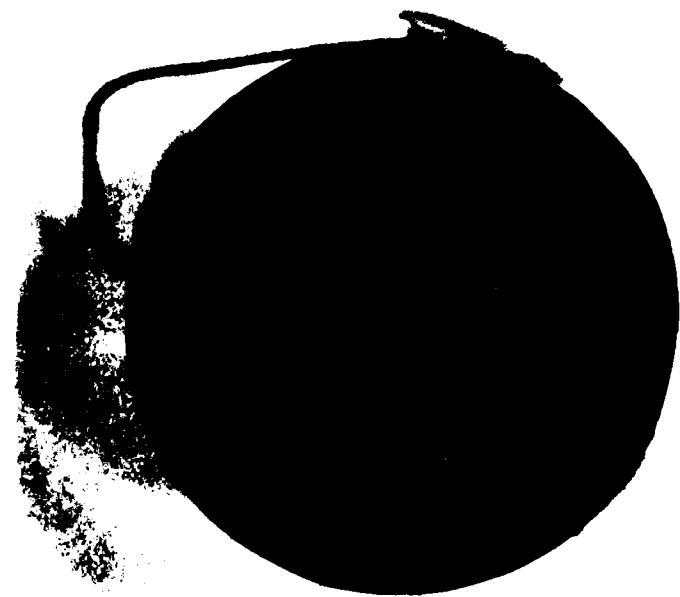


FIG. 11

CRACKING OF OVERLAY COATING WITH INTERVENING CHROMIUM LAYER

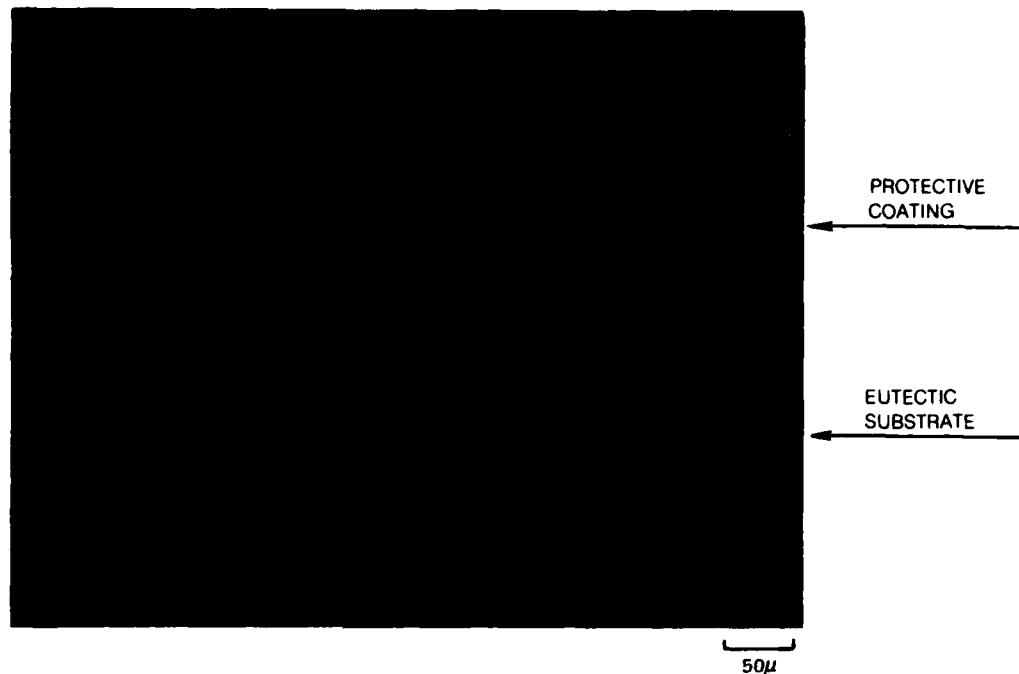


79-10-61-12

FIG. 12

FORMATION OF KIRKENDALL VOIDS AT COATING-SUBSTRATE INTERFACE

AS COATED MICROSTRUCTURE



AFTER ANNEALING TREATMENT (4 HOURS AT 1975°F IN ARGON)

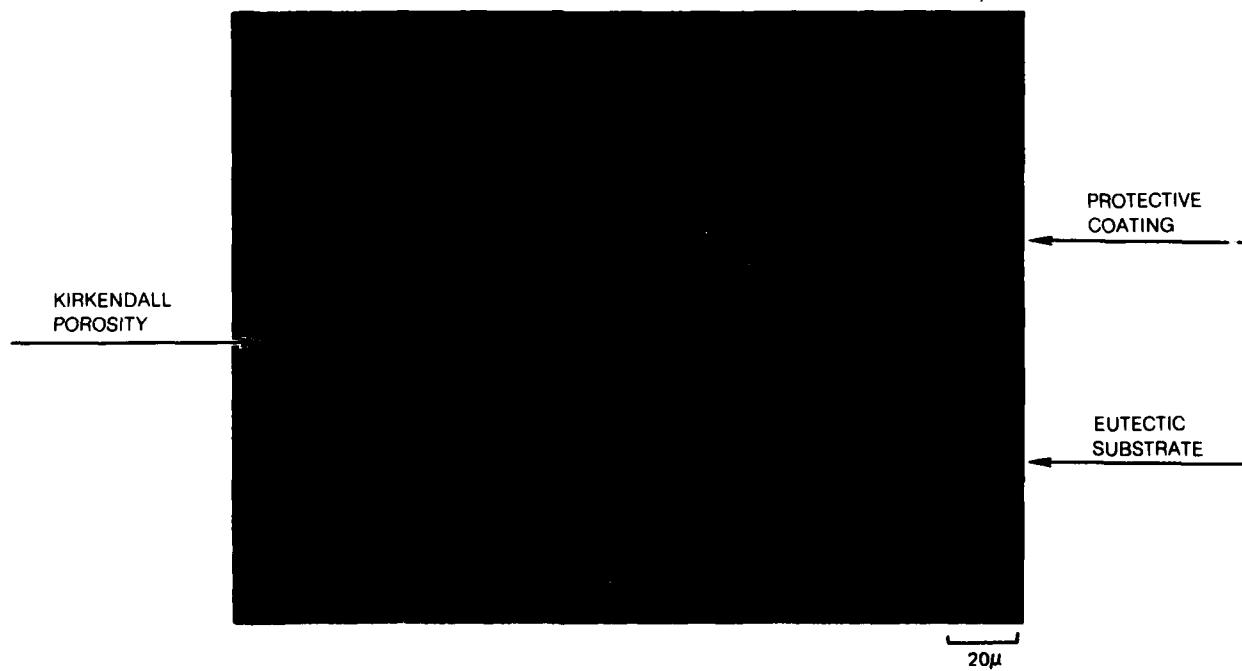
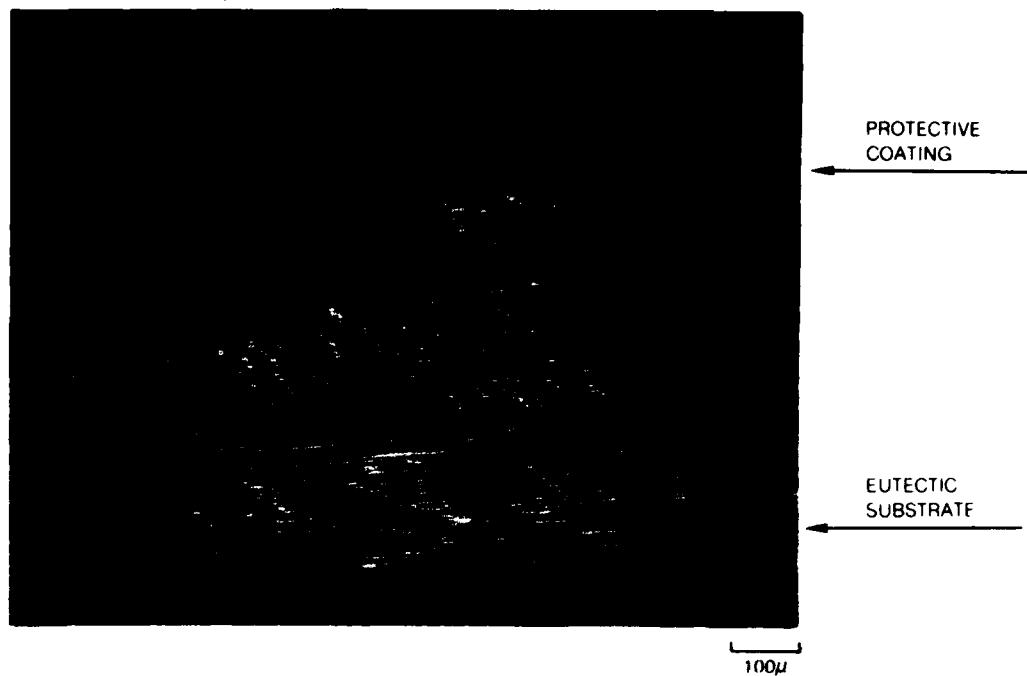


FIG. 13

CUT EDGE CORROSION

LONGITUDINAL SURFACE



TRANSVERSE SURFACE

